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THE EFFECT OF $CuAl_2$ DISPERSIONS ON THE FATIGUE OF ALUMINUM-COPPER ALLOYS

Twenty Sixth Technical Report

By

Robert B. Shaw

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April 6, 1953

Dr. O. T. Marzke
Chief, Metallurgy Section
Office of Naval Research
Department of the Navy
Washington 25, D. C.

Dear Dr. Marzke:

Attached hereto is a report on "The Effect of $CuAl_2$ Dispersions on the Fatigue of Aluminum-Copper Alloys" by Robert B. Shaw. This report is based on Mr. Shaw's M.S. Thesis investigation and was therefore supported by University funds. Since the subject matter for this thesis investigation was suggested by our earlier O.N.R. investigations on the effect of dispersions of $CuAl_2$ on the plastic properties of Al-Cu alloys, we wish to submit these data to O.N.R. in the form of the attached formal report.

Sincerely yours,

John E. Dorn

John E. Dorn
Professor of Physical Metallurgy

JED:bp

THE EFFECT OF $CuAl_2$ DISPERSIONS ON THE FATIGUE OF ALUMINUM-COPPER ALLOYS

By

Robert B. Shaw[(1)]

Twenty Sixth Technical Report, Series 22, Issue 26, N7-onr-295,

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April 1, 1953

(1) Graduate Student, Metallurgy Department, College of Engineering, University of California, Berkeley, California

ABSTRACT

The fatigue properties of Al-Cu alloys containing $CuAl_2$ dispersions were investigated at room temperature. In contrast to the well known results on spheroidal steels, the fatigue properties were found to be practically independent of the degree of dispersion. An investigation into the nature of crack formation in these alloys revealed that nucleation of microcracks depends primarily on the critical resolved shear stress on slip planes. Final failure was attributed to the accumulation of microcracks in the grain boundary regions.

INTRODUCTION

Our present concept of failure by fatigue is based on the idea of nucleation and growth of a crack. Upon repeated stressing, slip occurs in grains having suitable orientation with respect to the applied stress system. As cycling is continued, certain regions in these grains form cracks. These regions are believed to contain stress concentrations or imperfections, which either existed prior to stressing, or are subsequently developed. Guided by the state of stress and available paths of easy growth, a crack propagates through the polycrystalline metal until failure is complete.

Most investigations pertaining to the fatigue of metals fall into one of two general classes: those which deal with the effect of observable variables on the cycles to failure, and those which are concerned with the nucleation and growth of microcracks. Unfortunately, few investigations have been concerned with the fundamental mechanics of nucleation or growth of microcracks; the bulk of investigations have been concerned with the engineering aspects of fatigue. Inasmuch as we shall presently be concerned with both of these aspects of fatigue relative to aluminum-copper dispersion alloys, it would be well to review the results of these two classes of investigations from the standpoint of a comparison of single phase and two phase alloys.

It is a well known fact that low alloy steels exhibit endurance limits of about half of their tensile strengths. This may be associated with the homologous behavior of the mechanical properties of these steels. On the other hand, no similar trends have been found for two phase nonferrous alloys. The single phase alloys appear to behave normally, as is observed by the similarity between endurance ratios of binary ferrites and steels (1), and by the fact that the fatigue properties of aluminum alpha solid solutions are improved by increasing amounts of alloying element (2). Investigation of the fatigue properties of aluminum-copper dispersion alloys seemed warranted in view of the fact that the tensile properties of these dispersions have been shown to be homologous[3].

Nucleation of fatigue cracks in face centered cubic single crystals has been shown to be a direct consequence of the slip process, and thus related to the resolved shear stress on the slip plane[4]. Cracks invariably nucleate on slip planes, although they display a tendency to grow in the general direction of the maximum shear stress.

Nucleation of cracks in polycrystalline metals presents a more complicated problem. Evaluation of the state of stress on a microscopic scale is hampered by the presence of elastic anisotropy, plastic flow, and complex boundary conditions. Nevertheless, a qualitative picture is emerging from the investigations thus far completed. It has been observed that the general shapes of grains remain unaltered by fatigue. The grain boundaries, however, frequently become serrated as a result of slip on adjacent planes within a grain[5]. This would tend to support the concept that damage is localized to certain grains. Recent microscopic evidence has shown that a multitude of cracks nucleate and grow until they finally join together to form a number of large macroscopic cracks, one of which causes failure[6].

It appears that crack nucleation in polycrystalline metals may be governed by a shear stress law. Craig's electron microscopic studies on an aluminum - ½% silver solid solution alloy indicated that crack formation, as observed at 41,000X, is preceded by the appearance of a disturbed band, the crack subsequently opening in the center of this band[6]. These cracks were similar to those observed in single crystals at the usual low magnifications in that they were oriented parallel to slip traces.

There is some evidence which indicates that the mechanism of crack propagation differs in one and two phase aluminum alloys. In a single phase Al -½% Ag alloy, polygonization was observed at the head of the growing fatigue cracks, which resulted in the crack branching as it followed the new polygonized boundaries in its path[5]. This had the effect of slowing down the rate of crack growth. Recent x-ray evidence has shown that aluminum-silver solid solutions polygonize in both creep and tension while Al -2% Ag dispersions apparently do not polygonize[7]. These investigations are interesting because they suggest that the mechanism of crack nucleation and growth is strongly influenced by the presence of dispersions in aluminum solid solutions. Thus it would be desirable to conduct a microscopic investigation into the nature of crack formation in the Al-Cu dispersions, as well as to evaluate the effect of dispersions on the fatigue properties of these alloys.

MATERIALS FOR TEST

Alloys consisting of dispersions of $CuAl_2$ in aluminum alpha solid solution were selected as materials appropriate for this investigation. Their tensile properties are homologous and the relationship between microstructure and flow stress has been systematically investigated by Shaw et al[3]. In addition, the fatigue properties of Al-Cu alpha solid solutions are known[2].

Heat treatments were carefully selected so as to produce alloys with as wide a range of particle sizes as possible, compatible with ease of counting on photomicrographs. At the same time, care was taken to insure uniformity of grain size, distribution and size of precipitated particles, and a minimum of grain boundary precipitation or Widemanstatten effects. The chemical analysis of the alloys is given in Table I. The grain sizes and volumetric mean free paths of the alloys used are given in Table II. Calculation of mean free path through the continuous phase was made by using the equations derived by Shaw et al[3].

A general description of the heat treatments is as follows: after 72 hours homogenization at 540°C the alloys were cold rolled 30%, recrystallized at 540°C to obtain the proper grain size, and water quenched. The proper dispersion was achieved by precipitation at temperatures below the solvus, after which the

TABLE I

CHEMICAL COMPOSITION OF ALLOYS

Weight Percent Alloying Element

Chemical Analysis	Spectrochemical Analysis			
Cu	Si	Fe	Mn	Zn
3.05	.002	.002	Trace	.0004
4.03	.002	.002	.001	.0004
5.05	.003	.002	.002	.0006

TABLE II

GRAIN SIZES AND VOLUMETRIC MEAN FREE PATHS FOR DISPERSIONS OF $CuAl_2$

Dispersion	Grain Size* Grains / M.M.	Volumetric Mean** Free Path, Cm.
3-VC	0.8	.0097
4-C	2.4	.0064
5-C	2.3	.0076
3-F	2.4	.00098
4-F	2.3	.00078
5-F	2.1	.00063

* By intercept method

** The mean distance between particles of $CuAl_2$ in space. For the details of the calculation see reference (3).

material was held four days at 305°C to insure a uniform distribution of copper in solid solution. The heat treatments are given in Table III.

EXPERIMENTAL PROCEDURE

The investigation consisted of fatigue tests on the dispersion alloys followed by a microscopic investigation into the nature of the initiation of fatigue cracks.

The fatigue tests were conducted on a Krouse sheet type cantilever beam fatigue testing machine operated at 1750 cycles / minute. All tests were conducted at room temperature. Design of the specimen (as shown in Fig. 1) was such as to produce uniform longitudinal strain. Deflections at the apex induced equal cycles of positive and negative strain. The specimen thickness was .070 inches, achieved by cold rolling preceding the heat treatment. Homogenized sheet stock 0.100 inches thick was cold rolled 30%, heat treated to obtain the proper grain size and dispersion of $CuAl_2$, and finally held 4 days at 305°C. to stabilize the copper in the alpha solid solution. Heat treatments were conducted in a salt bath furnace to minimize surface oxidation. Fatigue specimens were subsequently machined with their axes in the rolling direction and their edges polished longitudinally with 3/0 emery paper.

Deflections were measured with a dial gauge mounted under the apex of the specimen. The Krouse Machine had been previously calibrated(2). The equation

$$\varepsilon = 0.1084\ D\ t$$

expresses the result of the calibration, giving the strain in the outer fiber as a function of the specimen thickness t, and the deflection at the apex D.

In the latter part of the investigation, the nature of microscopic fatigue cracks was studied. A large area of a 3-VC sheet fatigue specimen was electropolished on one side using a 50-50 mixture of nitric acid and ethyl alcohol as a

TABLE III

HEAT TREATMENTS FOR DISPERSION OF $CuAl_2$

Designation of Dispersion	Recrystallization Time Min.	Recrystallization Temp. °C.	Aging Treatment Time Hrs.	Aging Treatment Temp. °C	Aging Treatment Time Hrs.	Aging Treatment Temp. °C
3-VC	2½	540 WQ	24	465		
4-C	2	540 WQ	24	465		
5-C	4	540 WQ	1	525		
3-F	3	540 WQ	1/4	455 WQ	3	350
4-F	1	540 WQ	.025	415 WQ	3	350
5-F	2½	540 WQ	.025	525 WQ	1	400

Symbols: VC very coarse; C coarse; F fine; WQ water quenched.

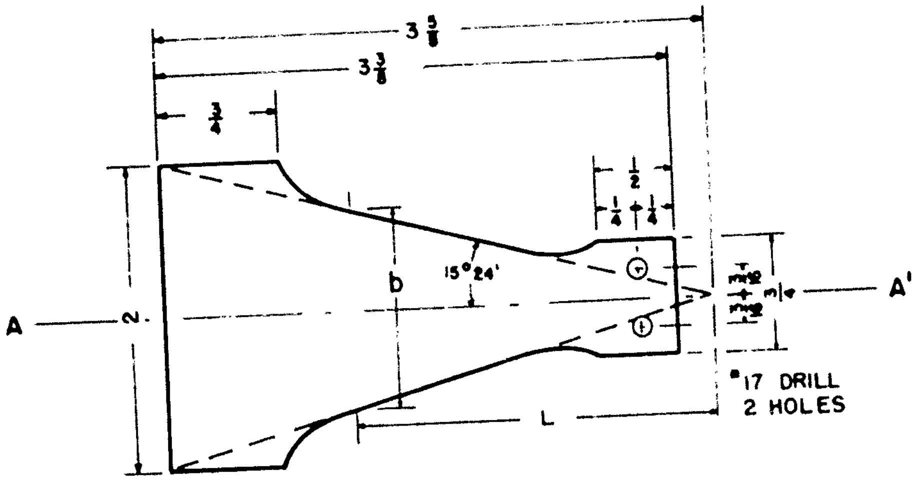

FIG. I FATIGUE SPECIMEN

polishing solution. An area covering about 50 grains was selected and mapped so that each grain could be located for future reference. Specific paths in each grain were selected and counts of the number of cracks traversed by the microscope in each grain were made at 600 X magnification after 800, 5,000 and 20,000 cycles. The lengths of the paths in each grain were measured using an eyepiece micrometer. The paths traversed by the microscope were in a direction perpendicular to the specimen axis (A - A' in Fig. 1).

The orientation of each grain was determined in the following manner: first, slip lines were generated by bending the specimen in the fatigue machine, and their angles with reference to the specimen axis were measured; next, the specimen was mounted on a goniometer head, and a grain centered with the aid of a microscope. The goniometer head was then placed in a diffractometer[8], which utilized a narrow beam of monochromatic copper radiation. The diffractometer could be preset at the proper Bragg angle for reflection of a given atomic plane, as the wave length of the radiation, and the interplanar spacing of the plane were known. Adjustment of the goniometer brought the specimen into the proper position for reflection, as indicated by the Geiger counter apparatus. The identity and location of the pole of a [111], [100] or [101] plane, in addition to the angles of two slip plane traces, provided sufficient information to determine the orientations. The orientations of forty-two grains were determined in this manner.

RESULTS OF FATIGUE TESTS

As aluminum alloys do not display a definite endurance limit, tests were terminated at 10^7 cycles if failure had not occurred. There was a tendency for fracture to start near the fillet region at the wide end of the specimen. Since the data correlated well as a whole, these data were included with the others.

Two alloys, 4-C and 4-F, were selected first for testing, as their mean free paths differed greatly, and would thus indicate the range of fatigue properties to be expected. These alloys also exhibited a considerable difference in particle size. As the results of Fig. 2 indicate, the change in fatigue properties was not large enough to warrant testing intermediate dispersions because the expected experimental scatter would be of the same order of magnitude as the differences in fatigue properties. Some of the other dispersions were tested at a strain of $\varepsilon = .001$. The results are shown in Table IV.

RESULTS OF MICROSCOPIC EXAMINATION

Preliminary examination of the electropolished surface of 4-C and 3-VC specimens, cycled at $\varepsilon = .001$, revealed to some extent the nature of the formation of the fatigue cracks. During the first 200 cycles, faint needle-like disturbances are observed parallel to the traces of (111) planes in a fraction of the grains. Upon additional cycling, these short faint lines darken and take on the aspect of short cracks (Figs. 3, 4 and 5). Some, but not all of the cracks nucleate at the edges of the $CuAl_2$ particles. New ones nucleate as the existing ones grow larger. The number and size of these "crack nuclei" vary from grain to grain. There is extensive nucleation and growth of cracks adjacent to grain boundaries coincident with crack formation within grains as shown in Fig. 6.

While the cracks within grains are growing at a slower rate, the grain boundary regions become thick, dark bands (Figs. 7 and 8). These bands are the result of the coalescence of a multitude of short cracks which have accumulated adjacent to either side of the boundary. Eventually these open into larger, visible cracks constituting about 80% of the path of the "final" crack. Thus, most of the "final" crack is made up of what may be termed "boundary cracking"; the remainder by the process of linking of short cracks within grains (Fig. 9).

The frequent appearance of crack traces oriented from 0° to 45° from the normal to the maximum principal stress, prompted an investigation of the effect of orientation on the frequency of appearance of cracks. A 3-VC specimen, one surface

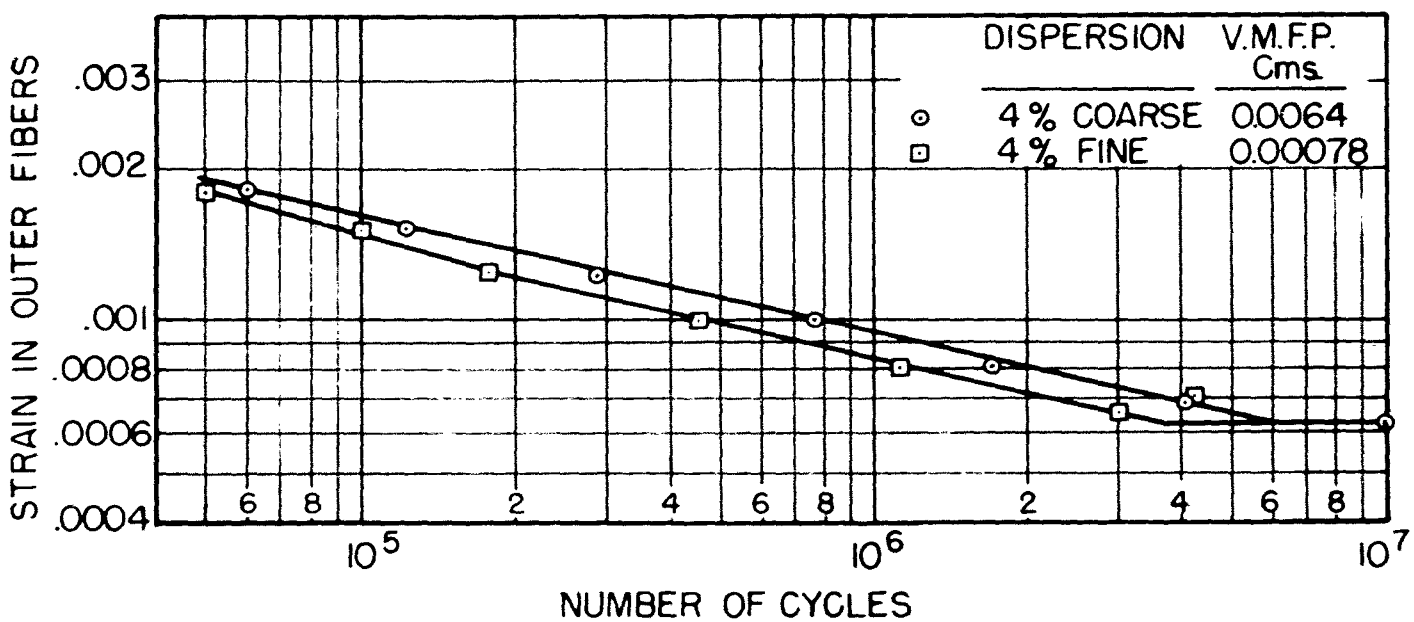

FIG. 2 THE EFFECT OF DISPERSION ON FATIGUE.

TABLE IV

COMPARISON OF FATIGUE STRENGTHS OF DISPERSION ALLOYS AT A STRAIN OF .001

Dispersion	V.M.F.P.* Centimeters	Cycles To Failure
5-F	.00063	4.1×10^5
4-F	.00078	4.5×10^5
3-F	.00098	4.8×10^5
5-C	.0076	5.5×10^5
4-C	.0064	7.5×10^5
3-VC	.0097	11.0×10^5

* The mean distance between particles of $CuAl_2$ in space. For the details of the calculation see reference (3).

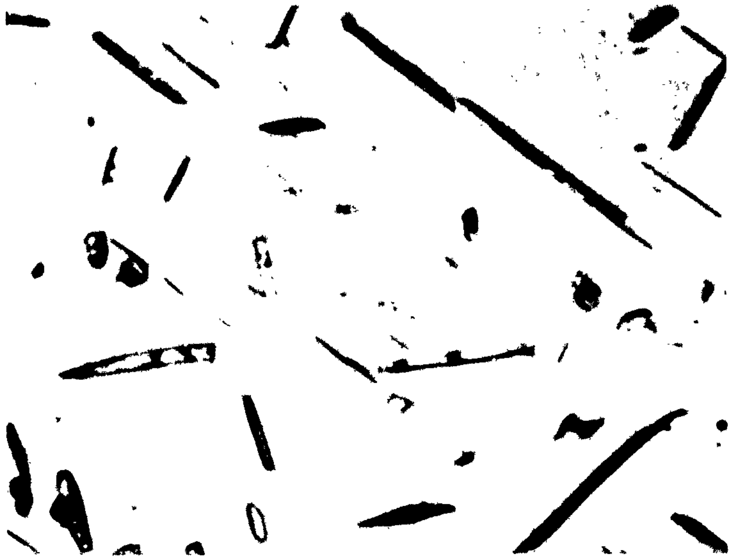

FIG. 3 5% COARSE SPECIMEN. MAGNIFICATION 600 X. 2000 CYCLES AT .001 STRAIN. NEEDLE-LIKE MICRO-CRACKS WITHIN A GRAIN.

FIG. 4 5% COARSE SPECIMEN. MAGNIFICATION 600 X. 2000 CYCLES AT .001 STRAIN. DUPLEX MICRO-CRACKS RELATED TO DUPLEX SLIP.

FIG. 5 5% COARSE SPECIMEN.
MAGNIFICATION 600 X.
20,000 CYCLES AT .001 STRAIN.
MULTIPLICATION AND GROWTH OF CRACKS WITHIN A GRAIN.

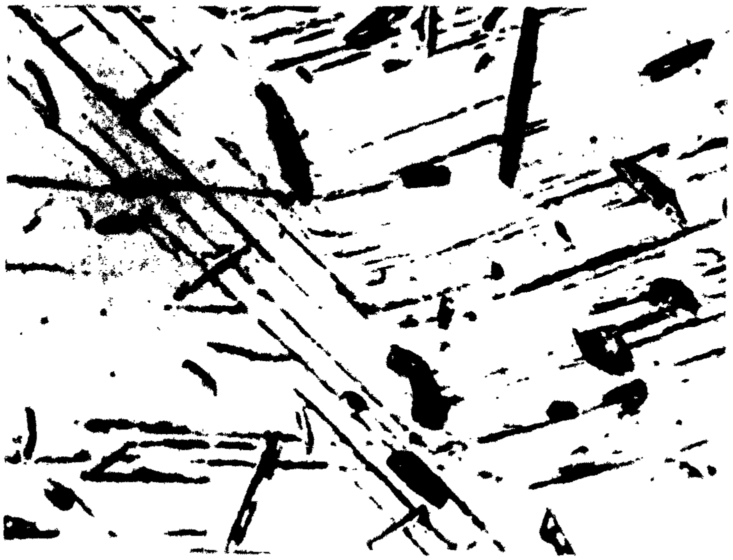

FIG. 6 5% COARSE SPECIMEN.
MAGNIFICATION 600 X.
20,000 CYCLES AT .001 STRAIN.
ACCUMULATION OF CRACKS IN THE VICINITY OF A GRAIN BOUNDARY.

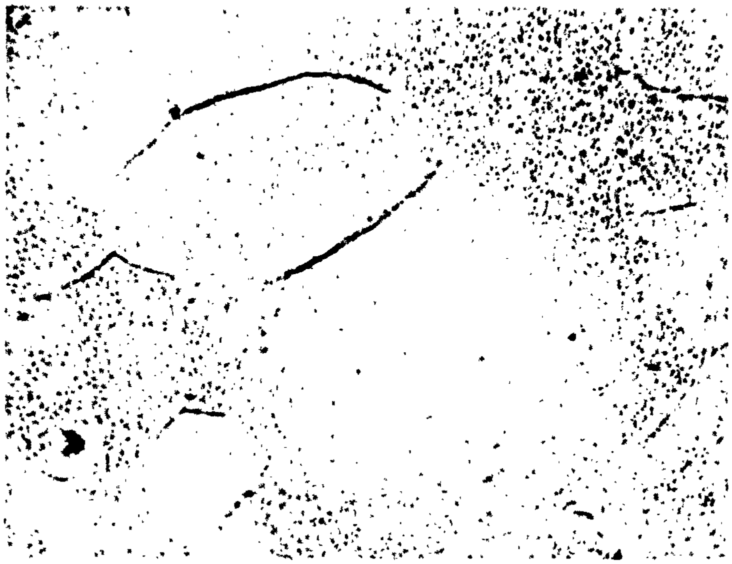

FIG. 7 3% VERY COARSE SPECIMEN.
MAGNIFICATION 50 X.
10,000 CYCLES AT .001 STRAIN.
PRINCIPAL STRESS IN THE VERTICAL DIRECTION. DARK BANDS IN GRAIN BOUNDARY REGIONS. SPECIMEN UNETCHED.

FIG. 8 3% VERY COARSE SPECIMEN.
THE SAME AS IN FIG. 7 AFTER 300,000 CYCLES.

FIG. 9 3% VERY COARSE SPECIMEN.
MAGNIFICATION 50 X.
300,000 CYCLES AT .001 STRAIN.
SPECIMEN UNETCHED. JOINING OF
SHORT CRACKS WITHIN GRAINS.

of which had been electropolished, was cycled at .001 strain. The number of cracks per millimeter after 800, 5000, and 20,000 cycles for specific paths in each grain, was measured at 600 X magnification. The grain orientations were determined and pole figures constructed. As the applied stresses are symmetrical about the longitudinal specimen axis, all [111] poles were plotted on one quadrant of the stereographic net. Figures 10, 11, and 12 show the effect of orientation of micro-cracks on their formation for 800, 5,000 and 20,000 cycles.

DISCUSSION OF RESULTS

The fatigue properties of Al-Cu dispersion alloys display surprising insensitivity to changes in the mean free path of the particles. Tenfold difference in mean free path in the 4% Cu alloy results in 25% increase in flow stress (at $\varepsilon = 0.20$), and virtually no change in the endurance limit. By comparison, the same dispersions in spheroidal steels exhibit 50% increase in flow stress, and about 50% increase in endurance limit. Apparently there is a lack of correspondence for dispersions in ferrite and dispersions in aluminum alpha solid solutions, regardless of the fact that both types of alloys exhibit homologous tensile properties. There appears to be no ad hoc reason why these two materials behave dissimilarly, although the following factors certainly influence the mechanism of failure:

(1) Crystal Structure

The effect of crystal structure on micro-crack formation is apparently related to the character and number of systems available for slip. A detailed analysis cannot be undertaken until the fundamentals of plastic deformation are more clearly understood. However, it may be concluded from the microscopic observations of Gough and Craig that there are distinct differences in the mechanism of crack formation in face centered cubic and body centered cubic crystals, and that these differences are related to the available slip systems.

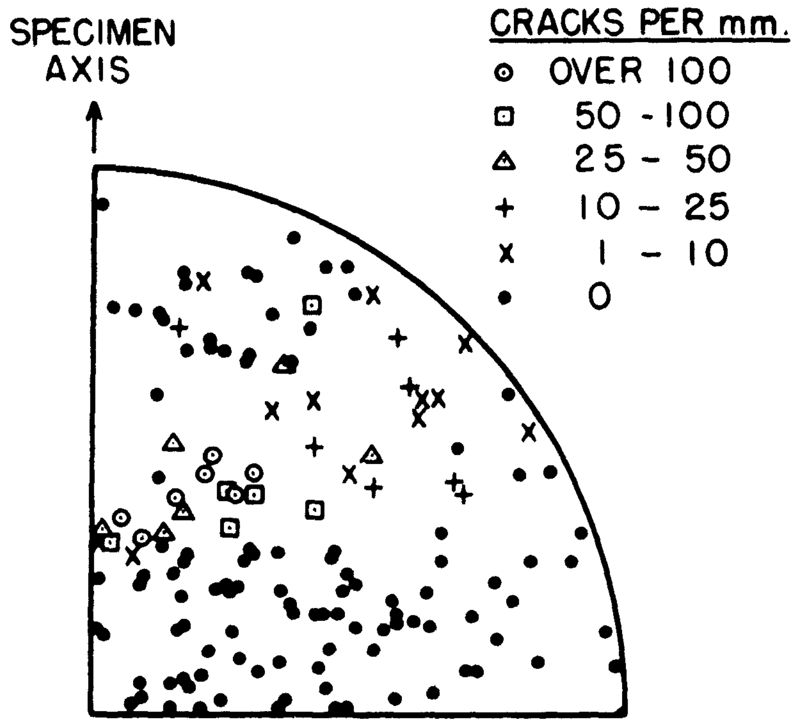

FIG. 10 THE EFFECT OF ORIENTATION ON THE FORMATION OF MICRO-CRACKS AFTER 800 CYCLES AT .001 STRAIN. POLES OF (111) PLANES ARE PLOTTED.

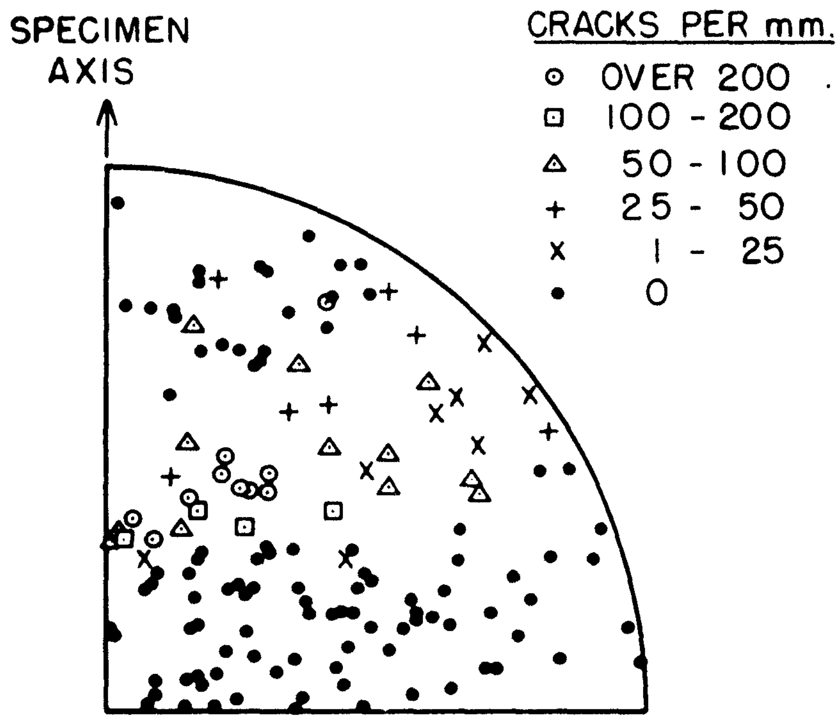

FIG. 11 THE EFFECT OF ORIENTATION ON THE FORMATION OF MICRO-CRACKS AFTER 5000 CYCLES AT .001 STRAIN. POLES OF (111) PLANES ARE PLOTTED.

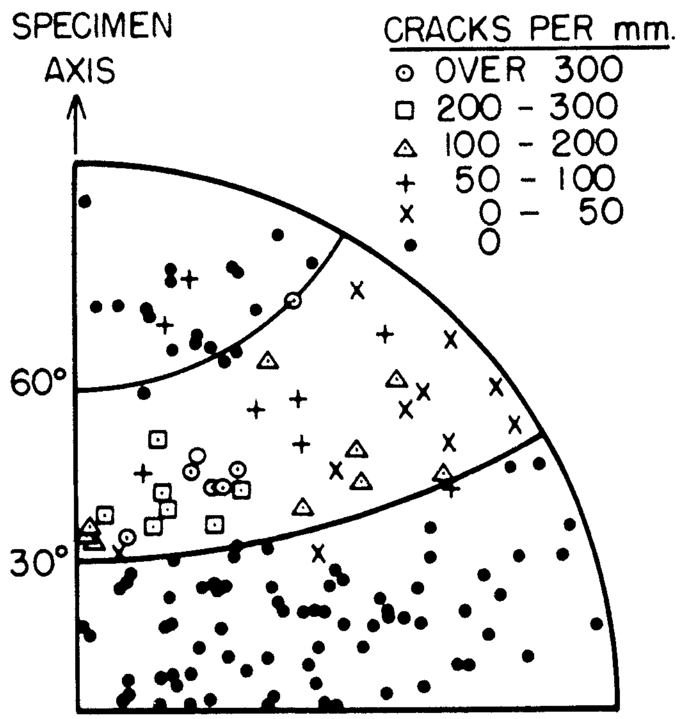

FIG. 12 THE EFFECT OF ORIENTATION ON THE FORMATION OF MICRO-CRACKS AFTER 20,000 CYCLES AT .001 STRAIN. POLES OF (111) PLANES ARE PLOTTED.

(2) Rates of Recovery at Room Temperature

Aluminum exhibits higher rates of recovery at room temperature than iron. Comparison of melting points, or Debye characteristic temperatures, for these metals show that low temperature tests on aluminum are, in a sense, equivalent to high temperature tests on iron, other factors being equal.

(3) Stress Concentrations Around Particles

The intensity of stress concentration around a hard particle depends to a marked degree on the shape of the particle. Spherodized cementite appears to be more nearly spherical than $CuAl_2$ (plate-like) particles, and would be expected to exhibit lower stress concentrations. In addition, the relative differences in the elastic properties of particle and matrix affects the stress concentration in the neighborhood of the particle. Stress concentrations undoubtedly appear at particles as a result of volume changes on precipitation, and differential contraction of particle and matrix upon cooling.

It is clear that the observed differences in fatigue properties of the Al-Cu dispersions and the spheroidal steels cannot be resolved simply in terms of differences in theoretical stress concentrations calculated on the assumption of a perfectly elastic matrix. For example, the stress concentration created by a perfectly rigid spherical particle in an elastic matrix subject to pure tension is approximately twice the background tension, and is independent of the size of the particle(9). This stress decreases to a low value at about two diameters from the center of the particle. If the volume fraction of dispersion is held constant, and the size of the particle is varied, the diameter of the particle and the distance between particles change in such a way as to maintain about the same degree of interaction regardless of particle size. Thus it might appear that the fatigue properties are independent of the dispersion; however, this conclusion must be discarded in view of the marked dependence of fatigue properties on dispersions in spheroidal steels.

The problem of fatigue failure is one of nucleation and growth of cracks and the influence of dispersions on these processes. The primary factor affecting nucleation in both single phase and in dispersion alloys appears to be the slip process, which is governed by the resolved shear stress law. In the fatigue of single phase aluminum alloys it appears that the process of polygonization at the head of the crack markedly affects subsequent crack growth, causing it to branch along the subgrain boundaries. In dispersion alloys this effect is not observed, the cracks being wholly crystallographic. Thus the effect of dispersions on crack propagation seems, in part, to be related to the effect of the dispersions on inhibiting polygonization. Tensile and creep data on aluminum alloys indicates that the presence of dispersions apparently inhibits the polygonization process[7]. Failure in the dispersion alloys occurs primarily as a result of accumulation of microcracks near grain boundaries, the cracking within grains being a secondary factor. The degree of damage to a particular boundary depends upon the orientations of the grains adjacent to that boundary and hence their cracking tendency. In addition, the orientation of the boundary itself seems to be important; boundaries lying near the normal to the principal tensile stress are most greatly affected. The role played by the individual particles in nucleation is uncertain, for if they serve as centers for nucleation one would expect the fatigue properties to be more markedly dependent on the dispersion. In these alloys the effect was small, and strangely enough, the coarser dispersions are slightly superior to the finer ones.

Results of the investigation on the effect of orientation of a crack on its formation are in good agreement with the findings of Gough on single crystals. The cracks appear to nucleate on those slip planes where the resolved shear stress is greater than some critical value. The pole figures exhibit a maximum crack nucleation in the central region on the plane of maximum shear stress. These results suggest that the fatigue properties of Al-Cu alloys could be improved by the introduction of preferred orientations in which the [111] poles fall outside the regions susceptible to rapid crack nucleation. The distribution of crack nucleation with respect to the applied stress system remains essentially unaltered as the number of cycles is increased.

CONCLUSIONS

1. The fatigue properties of Al-Cu dispersion alloys are practically independent of the degree of dispersion in the range investigated, the coarser dispersions exhibiting slightly higher fatigue strengths than the finer ones.
2. Nucleation of fatigue cracks is governed by a shear stress law. Only those slip planes which have a critical resolved shear stress or greater nucleate microcracks.
3. The immediate cause of failure is due to the accumulation and coalescence of crystallographically oriented microcracks in regions adjacent to grain boundaries, and their subsequent joining together with chains of short intercrystalline cracks.

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